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EFFECT OF HIGH HEAT INPUT ON THE MICROSTRUCTURE AND MECHANICAL PROPERTIES OF SUBMERGED-ARC WELDS IN STEELS FOR ARCTIC AND OFFSHORE STRUCTURES

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WELDING SECTION

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ABSTRACT

The microstructure/mechanical property relationships of the weld metal and heat-affected zone (HAZ) regions were evaluated for a series of submerged-arc welds deposited at 3 and 6 kJ/mm heat inputs in a range of C-Mn microalloyed steels. In addition, Gleeble simulation was employed to evaluate the HAZ notch toughness of the microalloyed steels.

The results showed that the HAZ notch toughness requirement (35 J at -40°C) could only be met using the Ti microalloyed, low carbon (350WT strand cast) steel. In general, this was attributed to the presence of stable (TiNb) (CN) precipitates which acted to restrict grain growth and produce a prior austenite grain size of approximately 55 μm . The remaining steels displayed poor toughness after simulation due to excessive grain growth resulting in a prior austenite grain size of the order of 120 μm .

The HAZ notch toughness of the real welds (measured 1 mm from the fusion line) exceeded the targeted level at the 3 kJ/mm heat input for all steels, however, only the 350WT strand cast steel exhibited adequate toughness at 6 kJ/mm. The loss of toughness at 6 kJ/mm was related to the increased proportion of HAZ being tested (size of local brittle zone) and to the overall coarsening of the microstructure.

The required weld metal notch toughness (35 J at -40°C) was only achieved for welds deposited at 3 kJ/mm using a multipass welding technique.

This resulted in a microstructure consisting of a relatively high proportion of acicular ferrite in the as-deposited regions, combined with approximately 20% reheated weld metal. In the 6 kJ/mm welds, the microstructure was coarser and a higher proportion of grain boundary ferrite contributed to the decrease in notch toughness.

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EFFET D'IMPORTANTES APPORTS DE CHALEUR, SUR LA MICROSTRUCTURE
ET LES PROPRIÉTÉS MÉCANIQUES DE SOUDURES EFFECTUÉES A L'ARC
SOUS FLUX, DANS DES ACIERS DESTINÉS A DES STRUCTURES ÉDIFIÉES
DANS L'ARCTIQUE ET DANS LES ZONES OFFSHORES

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RÉSUMÉ

On a évalué les relations entre la microstructure et les propriétés mécaniques du métal de soudure et des régions contenant des zones attaquées par la chaleur (HAZ) dans une série de soudures effectuées à l'arc sous flux avec des apports de chaleur de 3 et 6 kJ/mm, sur une gamme d'aciers microalliés C-Mn. En outre, on a employé la simulation de Gleeble pour évaluer la résilience d'entaille de la zone HAZ des aciers microalliés.

Les résultats ont indiqué que l'exigence concernant la résilience d'entaille de la zone HAZ (35 J à -40°C) ne pouvait être respectée qu'avec l'acier microallié Ti, à faible teneur de carbone (obtenu en coulée continue, 350 WT). En général, on a attribué ce détail à la présence de précipités stables (TiNb) (CN), qui exerçaient un effet limitatif sur la croissance des grains, et produisaient une granulométrie d'environ 55 µm pour l'austénite primaire. Les aciers restants montraient une faible résilience après la simulation, en raison d'une croissance excessive des grains, qui a eu pour conséquence une granulométrie de l'austénite primaire de l'ordre de 120 µm.

La résilience d'entaille des zones HAZ des soudures réelles (mesurée à 1 mm de la ligne de fusion) dépassait le niveau désiré en présence d'un apport thermique de 3 kJ/mm dans tous les aciers; toutefois, seuls les aciers 350 WT obtenus en coulée continue montraient une résilience suffisante à 6 kJ/mm. La perte de résilience à 6 kJ/mm était liée à l'accroissement de la proportion de zones HAZ testées (dimension de la zone fragile locale) et à l'accroissement global de la microstructure.

La résilience requise d'entaille des métaux de soudure (35 J à -40°C) n'a été atteinte que dans le cas des soudures déposées à 3 kJ/mm selon une technique de soudage en plusieurs passes. On a ainsi obtenu une microstructure composée d'une proportion relativement élevée de ferrite aciculaire dans les régions où le métal a été déposé à l'état fondu, combinée à approximativement 20% du métal de soudure réchauffé. Dans les soudures faites à 6 kJ/mm, la microstructure était plus grossière, et une plus forte proportion de la ferrite située à la limite des grains a contribué à la réduction de la résilience d'entaille.

INTRODUCTION

Over the past decade advances in steel making technology have permitted the use of high heat input submerged-arc welding (SAW) for joining Arctic grade steels while maintaining acceptable weldment properties. Several Japanese researchers (1-4) have published papers on the development of new steel processing methods and welding consumables designed specifically for high heat input welding. The HAZ properties of steels used mainly for Arctic icebreakers and offshore oil and gas production platforms have been investigated by a number of workers (5-7), the major emphasis being directed towards understanding microstructure/toughness relationships. Utilization of high heat input welding has generally been limited to the pipemill industry where studies have focussed on an evaluation of the properties of both the weld metal and heat-affected zone regions for submerged-arc weldments.

The anticipated design temperatures for Canada's east coast and Arctic are -30°C and -50°C respectively. In order to meet the property requirements at these temperatures it is necessary to develop steels with improved HAZ toughness and further improve welding consumables and procedures. Of interest to fabricators of Arctic and offshore structures is the evaluation of currently available Arctic grade microalloyed steels in order to establish their suitability for employing high-heat input SAW fabrication techniques.

It is well documented in the literature that adequate HAZ toughness can be attained at high-heat input by utilising Ti microalloy steels (10, 12-17). This is a result of the formation of stable TiN particles which pin the prior austenite grain boundaries thus inhibiting the rapid grain growth usually associated with the HAZ adjacent to the weld fusion line. Other factors that influence the toughness of the HAZ under high heat input welding conditions are:

- (a) A tendency towards the formation of coarse upper bainite microstructures,
- (b) Increased proportion of grain boundary ferrite,
- (c) Increased volume fraction and size of second phase regions, i.e., carbide particles, pearlite colonies and inclusions.

In addition to controlling grain coarsening by addition of Ti, the development of microalloy steels having good low temperature HAZ toughness at high-heat input has been made by reducing the size and distribution of harder phases by lowering the carbon equivalent.

In the present study the primary objective was to evaluate the effects of high heat input welding on the microstructure and resultant mechanical properties of submerged-arc welds deposited in a range of Arctic grade steels. In order to better differentiate the microstructure/property relationships of the HAZ of candidate steels a simulation procedure was used to isolate the CGHAZ microstructure over a larger volume for more meaningful toughness tests to be carried out.

EXPERIMENTAL

Base Materials

The chemical compositions of the base materials used in this experimental program are given in Table 1. Microalloy additions of Ni, Nb, V and Ti are present in various combinations. The EH36 steel has been used in the fabrication of offshore structures and Arctic vessels and was included as a reference composition (C-Mn-Nb) for evaluation and comparison purposes. The remaining steels (350WT CAB, LT60 and 350WT strand) were designed to meet the same nominal yield strength of 350 MPa, but also to provide improved low temperature toughness properties. Figure 1 and Table 2 show typical mechanical properties of the as-received base materials.

The steels were based on the following chemistries and processing techniques, EH36 (C-Mn-Nb) ingot cast and normalized, 350WT CAB (Mn-Ni-Nb-V) ingot cast controlled-rolled, calcium-treated, LT60 (C-Mn-Ni-Nb-V) ingot cast normalized, rare earth treated and 350WT strand (C-Mn-Ni-Nb-V-Ti) continuous cast and controlled-rolled.

Welding Procedure

Experimental welds were made in the steels listed in Table 1 using the submerged-arc welding (SAW) process. Two procedures were developed, the first a conventional multipass welding technique and the second a high-heat input two-pass welding technique. The multipass procedure employed a single electrode (4.0 mm diam.) at a nominal heat input of 3 kJ/mm. The joint design consisted of a 60° included angle bevel and a 6 mm land with no root opening.

The two-pass procedure utilized a tandem submerged-arc welding technique with a direct current electrode positive (DCEP) lead and AC trial electrode configuration. The joint design was modified to accommodate the increasing heat input (6 kJ/mm) by increasing the land to 12 mm with a 60°

included angle. The pass sequence and typical macrosections are shown in Fig. 2 and 3. All plates were prepared such that the welding direction was transverse to the plate rolling direction.

Both procedures employed carbon-arc backgouging on the second side prior to deposition of the final back pass or passes. All welding was completed using a wire/flux combination consisting of Oerlikon SD3/OP121TT. The welding conditions for each procedure are listed in Tables 3 and 4. All welds were inspected using gamma radiography.

Simulation of the CGHAZ

Standard Gleeble specimens (10x10x75 mm) were machined parallel to the rolling direction from each steel listed in Table 1. Weld thermal cycles were measured from thermocouples located in predrilled holes which coincided with the fusion boundary of a 3 kJ/mm bead-on-plate (BOP) weld. Data taken from the bay region of the 3 kJ/mm weld were chosen to represent the thermal history of the bay region of a single electrode 3 kJ/mm submerged-arc weld. The important parameters which made up the thermal cycle are as follows:

Peak Temperature	1375°C
Cooling Time $\Delta t_{800-500}$	20 s
Retention time above 1000°C	7 s

Mechanical Testing

Charpy-V-notch (CVN) specimens were machined from the experimental welds transverse to the weld length and notched through-thickness 1 mm from the fusion line position (Fig. 4), and at the weld centre line. Transition curves were produced by testing a minimum of three samples at key temperatures over the range -120 to +20°C.

Charpy specimens were prepared from all Gleeble cycled samples with a through-thickness notch centered within the simulated region and also tested to produce transition curves.

Vickers microhardness measurements were made at the weld centerline and in the bay region of the weldments and on simulated samples using a 500 g load.

Metallography

Sections taken from real weld and simulated specimens were prepared using standard metallographic techniques and subsequently etched in 2% nital. An estimate of prior austenite grain size was made for the real weld specimens in the bay region adjacent to the fusion line for both heat inputs and more accurate grain size measurements made on the simulated specimens using the circle intercept method. Microstructures at the root of the Charpy V-notches for selected real weld specimens were examined by metallographically preparing specimens after machining 2 mm from the surface.

Carbon extraction replicas were prepared from the bay region of real welds so as to include the fusion boundary and approximately 2 mm of HAZ. In addition, carbon replicas and thin foils were taken from the mid section of simulated specimens.

Examination of the replicas and the thin foils took place in a Philips TEM/STEM fitted with an X-ray energy dispersive analyser. Quantitative analyses of particles were made to determine the variation in composition with distance from the fusion line and how this compared with the simulated samples.

Following simulation a representative sample from each steel was notched and broken at liquid nitrogen temperature (-196°C) and examined in the scanning electron microscope to obtain an estimate of the cleavage facet size.

RESULTS

CGHAZ Microstructure of Simulated Specimens

Representative optical and SEM micrographs of the simulated CGHAZ structures are shown in Figs. 5-8. In all cases the general microstructure was typically bainitic with aligned and non-aligned ferrite with second phase or microconstituents in the form of either Fe_3C or M-A. The microstructure of EH36 in addition to the above contained a small volume fraction of martensite. A major difference was observed between the grain size of the 350WT strand (Fig. 8) and the other steels. The 350WT strand steel had an average prior austenite grain size of 55 μm while values for EH36, LT60 and 350WT CAB were 115 μm , 132 μm and 110 μm respectively.

Table 5 summarizes the data obtained from precipitate characterization of the base plate and simulated specimens. It can be seen that both LT60 and EH36 base material contained a bimodal distribution of particles with generally larger particles present in LT60. Following simulation of 3 kJ/mm the fine precipitates in LT60 dissolve leaving larger Nb-rich precipitates in the size range 100-400 nm. In contrast most of the precipitates in EH36 dissolve. The two Ti containing steels 350WT CAB treated and strand cast also contain bimodal distributions of precipitates in their base plate, however, the strand cast has a smaller mean size i.e., 77 nm compared with 150 nm for the CAB treated. In both steels the larger precipitates are (NbTi) (CN) with a Ti-rich core and Nb-rich "caps". After Gleeble thermal cycling all of the smaller particles and the Nb-rich caps dissolve leaving coarse particles with a more faceted morphology. It can be seen that the mean particle size in the stand-cast steel is again smaller than that in the CAB-treated steel.

An examination of thin foils revealed that the microstructure of all four steels was bainitic consisting of parallel laths of ferrite separated by low angle boundaries containing elongated plates of M-A constituent or carbides. It was identified that a colony of ferrite laths surrounded by high angle boundaries, usually called a bainite packet, made up the basic structural unit of these microstructures.

Hardness measurements taken from the centre of the simulated zone are presented in Table 6. It can be seen that similar values were obtained for 350WT CAB and LT60 steels and that 350WT strand cast was approximately 20-30 points below that of these steels. EH36 exhibited a much higher hardness of 323 DPH.

CGHAZ Microstructure of Real Welds

The real weld CGHAZ of all steels prepared at 3 kJ/mm displayed a coarser microstructure than their respective simulated specimens. In addition the real weld CGHAZ of both 350WT grades contained grain boundary ferrite whereas the corresponding simulated samples did not. Welds produced at 6 kJ/mm resulted in slightly coarser microstructures compared with those at 3 kJ/mm with an increase in volume fraction of grain boundary ferrite for the 350WT grades. The prior austenite grain size in the bay region adjacent to the fusion line was typically in the range 150-200 μm for all of the

steels welded at 3 kJ/mm and slightly larger, i.e., 200-250 μm for the 6 kJ/mm welds. Figures 9-16 show optical micrographs of the CGHAZ for all steels at both heat inputs.

Table 5 presents the results of precipitate characterization in the real welds. In all cases a region adjacent to the fusion line was devoid of precipitates typically 150 μm for LT60 and EH36 and 40-100 μm for the two 350WT grades. Beyond these distances the particle size and distribution varies depending on the steel. For example in LT60 at 3 kJ/mm coarse particles in the range 150-650 nm are present whereas EH36 has a finer distribution in the range 60-140 nm. These results are to be expected given the difference in precipitate size in the base materials of the two steels. The 350WT CAB treated steel welded at 3 kJ/mm and all of the 6 kJ/mm welds showed evidence of reprecipitation.

The microhardness values taken from the bay region of all welds adjacent to the fusion line are given in Table 6. It is evident that little change in hardness takes place on increasing heat input from 3 kJ/mm to 6 kJ/mm but there is a significantly lower hardness for the real welds compared with the Gleeble simulated specimens.

CGHAZ Mechanical Properties

The Charpy transition curves generated for the simulated and real weld CGHAZ regions are shown in Figs. 17-20. The 35 J transition temperatures are also summarized in Table 7.

The notch toughness results for the simulated CGHAZ were significantly lower than either the 3 or 6 kJ/mm real weld 1 mm HAZ data. There was little difference between the simulated HAZ toughness for steels 350WT CAB, EH36 and LT60 all of them exhibiting brittle behaviour over the complete test temperature range. In contrast, the 350WT strand cast steel displayed good toughness in the simulated condition (35J at -48°C). Scanning electron microscopy of fracture surfaces revealed that the average cleavage facet size for 350WT CAB, EH36 and LT60 was in the range 80-90 μm while for 350WT strand it was much lower at ~ 30 μm .

In all cases the 3 kJ/mm welds notched 1 mm from the fusion line (1 mm HAZ) exceeded the 35 J impact energy at -40°C . The weld MM4 deposited in 350WT strand steel was almost equivalent to the as-received base material. Ranking the steels in descending order of toughness for the 3 kJ/mm heat

input gives 350WT CAB followed by LT60 and EH36. The upper shelf energies relate to the residual element levels in each steel. In general higher shelf energy was recorded in the steels with the lowest residual sulphur contents, with 350WT strand cast steel exhibiting the highest energies.

Weld Metal Chemistry and Microstructure

The chemical composition of the submerged-arc welds are listed in Tables 8 and 9. The same wire/flux combination was used for all procedures, therefore, only slight variation in C, Si and residual elements, i.e., Nb, V, and Ti occurred for a given heat input. The variations were attributed principally to dilution from the various base materials and/or welding wire chemistry. In general, there was an increase in the residual element content with increasing heat input. Additionally, it should be noted that the oxygen content tended to decrease at the higher (6 kJ/mm) heat input.

Metallographic examination of the two series of welds revealed significant differences in their microstructures. The welds deposited at 3 kJ/mm were composed of as-deposited, coarse and fine-grained reheated weld metal regions, whereas the 6 kJ/mm welds contained only an as-deposited region in the location of the Charpy specimens. There was approximately 20% reheated weld metal within the Charpy region of the 3 kJ/mm welds.

Representative micrographs of the as-deposited columnar structure for the 3 kJ/mm welds are shown in Fig. 21. Figure 21(a) shows the columnar structure in which relatively large prior austenite grains are surrounded by grain boundary ferrite and ferrite-with-second phase. Acicular ferrite predominates within the prior austenite grains (Fig. 21(b)). The coarse grained reheated microstructure was similar to that of the as-deposited region in that grain boundary ferrite and intragranular acicular ferrite were present, however, the prior austenite grains tended to be more equiaxed as shown in Fig. 22(a). In the fine-grain reheated region the microstructure consisted of fine polygonal ferrite and a carbide aggregate (pearlite and/or bainite), shown typically in Fig. 22(b). For the 6 kJ/mm welds which contained only as-deposited weld metal in the vicinity of the Charpy specimens the columnar structure tended to be essentially vertical at the weld centerline (Fig. 23). The prior austenite grains were narrower and surrounded by coarse grain boundary ferrite. Coarser acicular ferrite and polygonal ferrite were observed within the grains Fig. 23(b).

Mechanical Properties

The tensile properties of the submerged-arc welds are presented in Table 10. All welds exceeded the manufacturers specifications for

Yield Strength	427 MPa
Ultimate Strength	538 MPa
Elongation	25 %

Variations in yield strength existed between welds at a given heat input despite the fact that the same welding procedures and consumables were used. These variations could not be satisfactorily explained on the basis of the chemistry and resultant microstructure.

Charpy-impact transition curves generated for the weld metals are shown in Fig. 24. All the welds deposited at 3 kJ/mm met the targeted impact requirement of 35 J at -40°C with MM4 exhibiting the highest toughness.

In all cases the 6 kJ/mm welds failed to meet the minimum 35 J impact energy at -40°C as shown in Table 7. The transition curves were generally shifted to higher temperatures by 40°C to 50°C (Fig. 24(b)).

DISCUSSION

Heat-Affected Zone

Simulation has proven useful in identifying microstructure-toughness relationships in the HAZ (10, 12, 16, 18, 19) although there are limitations to the accuracy with which the microstructure of the real weld CGHAZ (grain size, precipitation and microconstituent) can be simulated exactly. For example, it has been documented that when the thermal cycle typical of the HAZ adjacent to the fusion line is reproduced in a thermal simulator the grain size of the simulated specimen is larger than that in the real weld HAZ particularly at low heat input. This is due to the steep peak temperature thermal gradient present in the real weld restricting grain growth.

A comparison of the real weld and simulated HAZ microstructures, however, revealed that the real weld HAZ grain size of all four steels was slightly larger than that in the simulated material. This difference was attributed to the variation in bay region profiles between the bead-on-plate weld from which the thermal cycle was measured and the final pass of the 3 kJ/mm multipass weld examined in this study. This is readily explained by

the fact that the measured thermal cycle was taken from a 3 kJ/mm bead-on-plate weld (time above 1000°C = 7 s) whereas the actual multipass weld at this heat input had an interpass temperature of 150°C, thus increasing the cooling time and time spent above 1000°C almost by a factor of two.

The notch toughness results (Table 7) for the simulated CGHAZ showed that the 350WT strand cast steel exhibited the best overall toughness compared with the other steels. Because of the significant decrease in toughness after the 3 kJ/mm simulation no tests were performed at a 6 kJ/mm heat input. The superior toughness of the 350WT strand-cast steel was attributed to its much smaller grain size brought about by the presence of fine (TiNb) (CN) cuboidal precipitates. It has been reported that these particles remain stable during the weld thermal cycle up to temperatures approaching those experienced adjacent to the fusion line. It has been shown, however, that their stability is a function of size, composition and peak temperature (21-23). In 350WT strand a sufficient volume fraction of particles in the size range 10-70 nm is present which acts to restrict the grain growth and produces a prior austenite grain size at least half that of the remaining steels. The latter exhibit either significant coarsening of precipitates or complete dissolution. Examination of the fracture surface reveals that the cleavage facets are less than the prior austenite grain size which is consistent with the results of various studies on the properties of microalloy steels containing upper bainite microstructures in the HAZ (12,24,25) (see Fig. 25). These studies have shown that a unit smaller than the prior austenite grain size commonly termed a "bainite packet" is surrounded by high angle boundaries and plays a major role in determining the toughness. The notch toughness of 350WT strand is further improved by having a lower carbon content and thus a lower volume fraction of carbides. It is evident from Fig. 14 that 350WT strand, apart from having a lower transition temperature, has a much higher upper shelf compared with the other steels arising from a lower level of residual elements and consequently a lower volume fraction of inclusions. To summarize, the effect of imposing a 3 kJ/mm simulation on steels LT60, 350WT CAB and EH36 was to produce a significant grain coarsening with a dramatic loss with toughness compared with the base material (Table 7). The 350WT strand cast steel, however, due to the presence of a fine distribution of (TiNb) (CN) which restricted grain growth together with lower C and residual element levels compared to the other steels displayed far superior toughness.

In general, the HAZ notch toughness results for the real welds show that there was a trend towards decreasing impact energy with increasing heat input. Although the HAZ of the real welds deposited at 3 kJ/mm exceeded the targeted impact energy requirement of 35 J at -40°C only the HAZ tests conducted for the 350WT strand cast steel exhibited adequate toughness at 6 kJ/mm (Table 7). The relatively high toughness exhibited at 3 kJ/mm and displacement of the Charpy transition curves to higher temperatures at 6 kJ/mm (Figs. 17-20) was attributable to the following two factors; (1) on a macroscopic scale, the increased heat input resulted in a larger proportion of CGHAZ being sampled which tends to lower the toughness and (2) an overall coarsening of the HAZ microstructure due to the slower cooling rate of the high-heat input welds.

Many studies have been published showing that macroscopic heterogeneity or local brittle zones affect the toughness in composite test specimens, i.e., specimens containing weld metal, CGHAZ, and base metal (6, 26-31). The work of Satoh et al. in particular (28,30), has shown this for CTOD and Charpy testing. They demonstrated that as the proportion of embrittled region increases the toughness approaches that of the lowest toughness component. In the present study the proportion of CGHAZ was estimated to be $\sim 20\%$ for the 3 kJ/mm welds whereas the 6 kJ/mm welds contained $\sim 50\%$ CGHAZ. Figure 26 shows the fracture path for weld MM8 corresponding to the 6 kJ/mm high-heat input welds. It can be seen that the crack front closely followed the fusion line. These observations are in agreement with those of McGrath et al. (29) who found that the HAZ notch toughness was only affected at higher heat inputs $= > 3$ kJ/mm in HY80 weldments where the proportion of embrittled CGHAZ represented 30 to 50%. Further evidence for the effect of local brittle zones has been demonstrated by Kudoh and Pisarski (31) who introduced local brittle zones into C-Mn multipass welds and found that when 10% of the Charpy notch tip sampled low toughness weld metal the notch toughness was unaffected. However, when the embrittled region was increased to 30% the toughness was reduced to that of the lowest toughness component sampled by the notch tip.

In addition, differences in microstructure observed for the 6 kJ/mm welds (Figs. 9 and 10) compared to the 3 kJ/mm welds will contribute to the decrease in toughness. These changes consist of a coarsening of the prior austenite grain size together with an overall coarsening of the general microstructure.

Weld Metal

The results from weld metal notch toughness testing of the 3 and 6 kJ/mm welds have shown that although the specified tensile properties (Table 10) could be met, the required impact energy (Table 7) could only be achieved using the 3 kJ/mm multipass welding technique with the consumables employed in this study. The results can be explained in terms of the effects of the weld metal cooling rate and chemistry on the microstructure and its resultant mechanical properties. Generally, the final weld microstructure developed for a given welding condition is considered dependent on the chemical composition, cooling rate, austenite grain size and inclusion content (32,33).

The superior toughness exhibited by the 3 kJ/mm welds in this work was attributed to the fine acicular ferrite structure combined with approximately 20% reheated weld metal and low inclusion content (O+S). Carbon, manganese and silicon provided sufficient weld metal hardenability to allow a relatively high proportion (>50%) of acicular ferrite to be formed at this heat input and inclusion content. It is generally accepted that acicular ferrite offers the best resistance to cleavage fracture by virtue of its fine unit grain size typically 1 to 3 μm wide (32-34). However, the work of Tweed and Knott (34) and Evans (35,36) has also shown that increasing the proportion of reheated weld metal may impart better toughness than as-deposited weld metal which contains coarse grain boundary ferrite. Several authors support the view that a decrease in the proportion of grain boundary ferrite will enhance as-deposited weld metal toughness, particularly for high heat input welds, (11,32,37).

In the case of the 6 kJ/mm two pass welds the slower cooling rate $\Delta t_{800-500} \sim 60$ s resulted in the formation of coarser acicular ferrite and grain boundary ferrite compared with the 3 kJ/mm welds (Figs. 21-23). It can be stated that for this weld metal composition an increase in heat input results in higher transformation temperatures and a coarsening of the microstructure. In turn this produces the observed shift in the Charpy transition curves to higher temperatures. The higher heat input produces a greater dilution of V and similar level of Nb compared with the 3 kJ/mm welds but from previous studies (38,39) it appears that these levels are not sufficient to have a detrimental effect on the toughness.

It can be seen from the notch toughness data that there were differences in impact energy between the steels at both heat inputs in the ductile region. These differences are most likely due to variations in O+S levels and could only be clarified by testing at higher temperatures.

In order to improve the toughness levels at high heat inputs a weld metal microstructure containing a high proportion of acicular ferrite with minimal grain boundary ferrite is desirable. This may be achieved by employing an alloyed electrode with increased hardenability. Several workers (8,11,32,40-43) have shown that Ti-B, Mo and/or Ni electrodes may be suitable for high heat input welding, however, all welding procedures would require procedure qualification testing for each base material composition and wire/flux combination.

CONCLUSIONS

1. Of the four microalloy steels simulated at a heat input corresponding to 3 kJ/mm only 350WT strand exhibited a notch toughness which exceeded 35 J at -40°C.
2. The superior notch toughness displayed by 350WT strand in the simulated condition was attributed to a much finer prior austenite grain size compared with the remaining three steels. This finer grain size was due to the presence of a fine distribution of stable (TiNb) (CN) precipitates which acted to inhibit grain growth during the weld thermal cycle.
3. The real weld HAZ notch toughness (measured 1 mm from the fusion line) exceeded 35 J at -40°C for all four steels for welds deposited at 3 kJ/mm heat input and only the 350WT strand cast steel exhibited adequate toughness at 6 kJ/mm.
4. In general, the HAZ notch toughness of the real weld was controlled by the following two factors: (1) the proportion of HAZ taking part in the fracture (size of local brittle zone) and (2) a coarsening of the overall microstructure.
5. The weld metal tensile (yield strength ≥ 427 MPa) and notch toughness (35 J at -40°C) property requirements were achieved for the metals deposited at 3 kJ/mm heat input. The toughness was attributed to a relatively high proportion of acicular ferrite combined with approximately 20% reheated weld metal.

6. Although the tensile properties were met at the 6 kJ/mm heat input, the notch toughness was substantially reduced due to coarsening of the microstructure and an increased proportion of grain boundary ferrite.

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Table 1 - Chemical composition of base materials

Base Material	Element wt %												
	C	Mn	Si	S	P	Al	Ni	Cr	Mo	Ti	V	Nb	Cu
350WT CAB	0.15	1.44	0.240	0.005	0.015	0.015	0.225	0.240	---	0.007	0.060	0.020	0.015
EH-36	0.14	1.48	0.255	0.025	0.025	0.040	0.020	0.200	0.040	0.005	<0.005	0.031	0.025
LT-60	0.14	1.35	0.280	0.010	0.015	0.030	0.125	0.175	<0.005	0.005	0.053	0.039	0.180
350WT strand	0.07	1.37	0.280	0.005	0.005	0.040	0.225	0.270	<0.005	0.010	0.068	0.022	0.015
													0.38
													0.17

Note: Sn <0.005

Table 2 - Tensile properties of as-received base materials

Base Material	Yield strength (MPa)	Ultimate strength (MPa)	Elongation (%)	Reduction in area (%)
350WT CAB	496	600	32	72
EH-36	407	545	33	71
LT-60	414	530	36	76
350WT strand	407	496	36	77

Table 3 - Welding parameters used for 3 kJ/mm welds

Side No.	Pass No.	Wire Feed Speed (mm/s)	Current (A)	Voltage (V)	Travel Speed (mm/s)	Electrical Stick Out (mm)
1	1-5	21.1	~600	32	6.4	32
2	6,7	21.1	~600	32	6.4	32

Table 4 - Welding parameters used for 6 kJ/mm welds

Side No.	Pass No.	Current (A)	Voltage (V)	Travel Speed (mm/s)	Electrical Stick Out (mm)
1	1	1000 (DC) 800 (AC)	34 38	10.5	35
2	2	1000 (DC) 800 (AC)	32 38	10.5	35

Table 5 - Summary of precipitate composition, size and distribution

	350WT CAB	EH36	LT60	350WT Strand
Base Metal	bimodal 1. 50-450 nm Ti-rich core Nb caps 2. <20µm higher Nb some V + Ti	bimodal 1. 50-300 nm 2. 5-15 nm	bimodal 1. large Nb-rich 125-650 nm 2. smaller Nb+V <50 nm	variety of particle morphologies 1. coarse ppts - faceted or rounded cuboids 2. fine spheroidal
3 kJ/mm sim.	Nb rich caps dissolve, small particles dissolve, faceted ppts remain av. size 150 nm	few large mostly dissolved	fine ppts dissolved larger particles remain 100-400 nm	fine spheroidal dissolve ppts become more faceted with increase in Ti/Nb ratio
6 kJ/mm sim.	Nb rich caps dissolve small particles dissolve av. size of remaining 90 nm	as above	as above	similar to base metal, coarse ppts show increase in Ti/Nb ratio
3 kJ/mm weld	<0.1 mm from FL no ppts. then coarse ppts, in addition fine scale (5-20 nm) ppts	<0.15mm from FL no ppts 0.15mm-1mm from FL size from 60 nm-140 nm then no change	<0.15mm no ppts coarse ppts away from fusion line 150-650 nm	40-80 µm from FL devoid of ppts 0.1mm from fusion line (10-30 nm) 0.5mm range of particle sizes
6 kJ/mm weld	<0.1 mm from FL no ppts then coarse ppts, also fine scale ppts 5-20 nm	similar to above but with reprecipitation	<0.15 mm no ppts coarse ppts, away from FL 150-650 nm also reprecipitation	40-80 µm from FL is devoid of ppts. similar to above

Table 6 - Microhardness values of simulated and real weld CGHAZ

Steel	3 kJ/mm simulated	3 kJ/mm real weld	6 kJ/mm real weld
350WT CAB	256	225	214
EH-36	323	235	228
LT-60	267	219	219
350WT strand	239	210	209

Table 7 - Charpy impact 35J T.T, °C

Base Material	as-received	3 kJ/mm simulated CGHAZ	3 kJ/mm real weld CGHAZ	6 kJ/mm real weld CGHAZ	3 kJ/mm weld metal	6 kJ/mm weld metal
350WT CAB	<-80	0	<-80	-25	-50	+12
EH36	-50	+13	-60	-25	-50	-10
LT60	-105	- 7	-60	-32	-50	+ 8
350WT strand	-120	-48	<-100	-72	-60	-10

Table 8 - Chemical composition of 3 kJ/mm submerged-arc weld metals

Weld No	Base Material	Element wt %												
		C	Mn	Si	S	P	Ni	Cr	Mo	V	Nb	Cu	N	O
MM1	350WT CAB	0.12	1.4	0.29	0.006	0.015	0.14	0.10	---	0.020	0.014	0.12	0.0087	0.0254
MM2	EH36	0.10	1.4	0.31	0.010	0.020	0.04	0.07	<.005	0.005	0.014	0.09	0.0077	0.0284
MM3	LT-60	0.14	1.4	0.30	0.007	0.016	0.11	0.14	<.005	0.015	0.015	0.17	0.0084	0.0304
MM4	350WT strand	0.10	1.4	0.33	0.006	0.016	0.15	0.17	<.005	0.025	0.015	0.14	0.0076	0.0225

Note: For all welds; Sn = 0.015

Al = 0.020

Ti = 0.008

Table 9 - Chemical composition of 6 kJ/mm submerged-arc weld metals

Weld No	Base Material	Element wt %														
		C	Mn	Si	S	P	Al	Ni	Cr	Mo	Ti	V	Nb	Cu	N	O
MM5	350WT CAB	0.12	1.4	0.280	0.005	0.012	0.010	0.110	0.06	---	0.010	0.025	0.015	0.07	0.0098	0.0183
MM6	EH36	0.15	1.4	0.290	0.015	0.018	0.020	0.075	0.12	---	0.008	0.005	0.020	0.14	0.0068	0.0159
MM7	LT-60	0.13	1.4	0.290	0.009	0.013	0.025	0.115	0.17	0.010	0.010	0.030	0.015	0.07	0.0081	0.0196
MM8	350WT strand	0.09	1.4	0.280	0.006	0.008	0.020	0.025	0.13	0.025	0.010	0.030	0.015	0.07	0.0075	0.0164

Note: For all welds; Sn = 0.010

Table 10 - Tensile properties of submerged-arc welds

Weld No	Yield strength MPa	Ultimate strength MPa	Elongation %	Reduction of area %
MM1	505	619	31	66
MM2	541	629	29	67
MM3	538	635	27	64
MM4	582	653	26	68
MM5	548	669	27	64
MM6	535	692	26	59
MM7	542	670	25	64
MM8	498	634	25	64

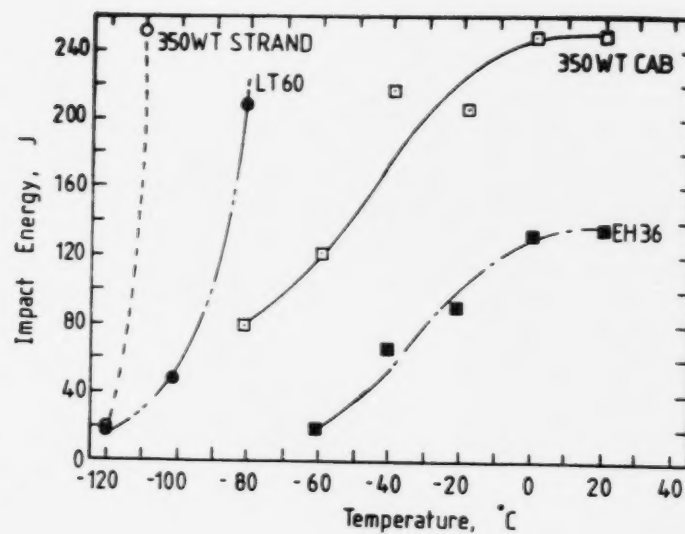


Fig. 1 Charpy transition curves of as-received base materials. Specimens were machined parallel to the rolling direction and notched through-thickness (neg. 15477).

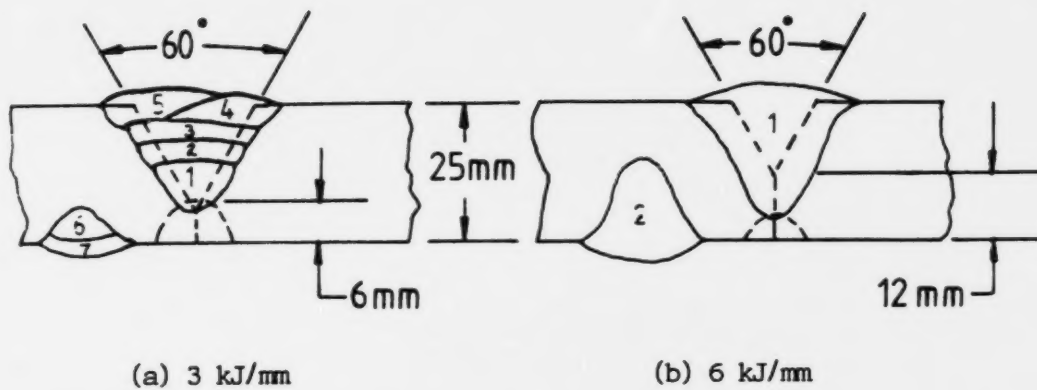
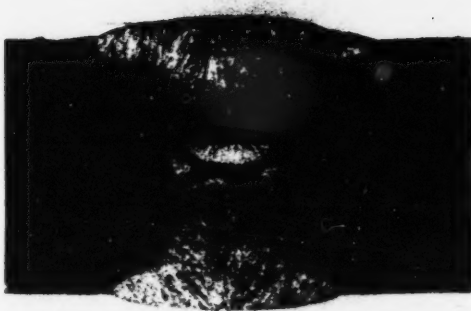
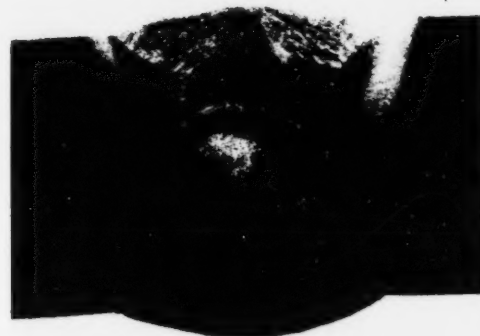


Fig. 2. Typical joint design and pass sequence for submerged-arc welds (neg.15488).



(a) 3 kJ/mm (neg.17339)



(b) 6 kJ/mm (neg.17340)

Fig. 3. Representative macrosections of submerged-arc welds showing pass sequence.

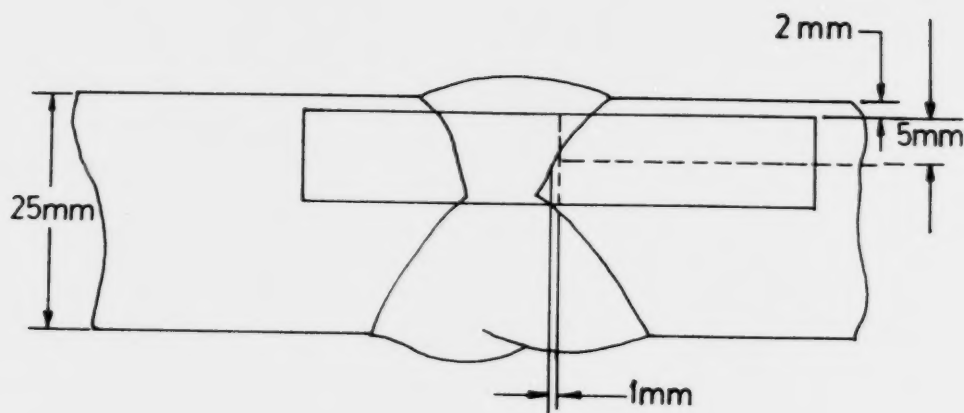
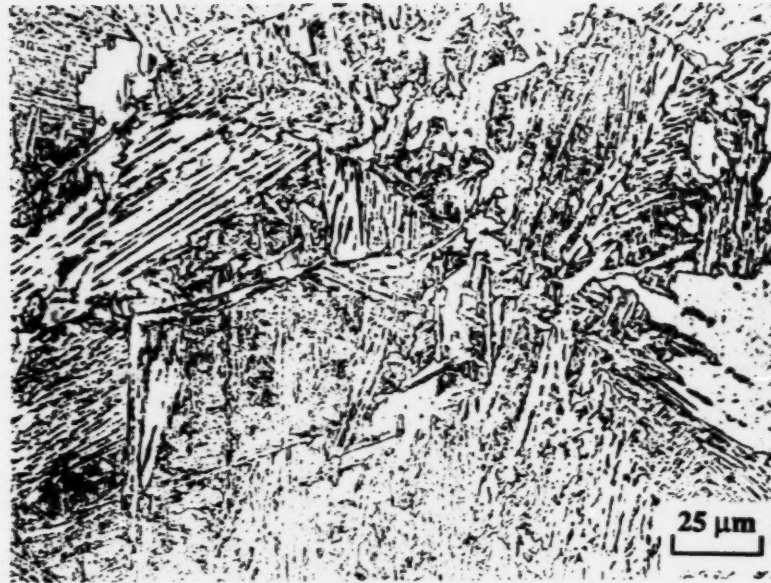


Fig. 4. Schematic diagram showing HAZ Charpy location 1 mm from the fusion line (neg. 14093).

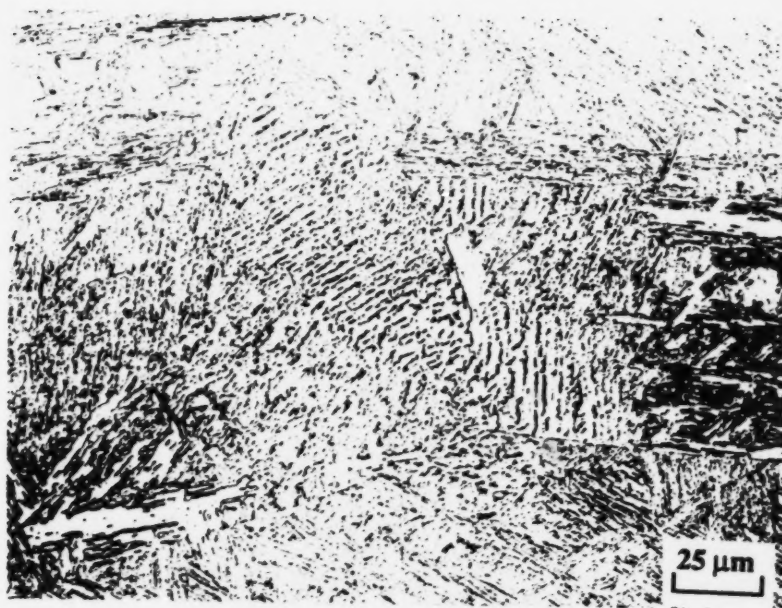


(a) Optical (neg.17341)

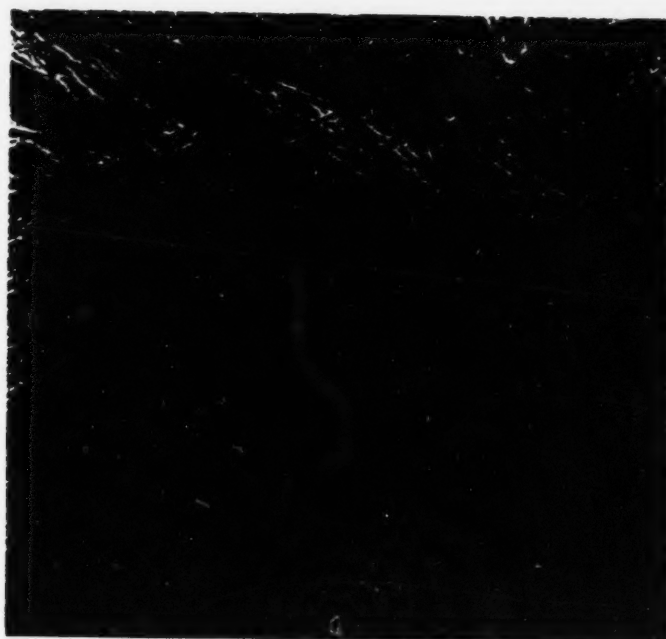


(b) SEM (neg.35 mm)

Fig. 5. Optical and SEM micrographs of 3 kJ/mm simulated CGHAZ microstructure in 350WT CAB steel.

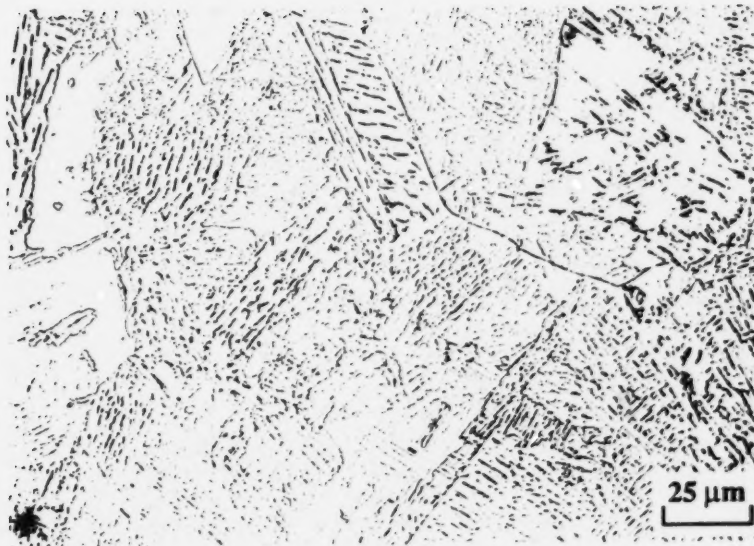


(a) Optical (neg.17342)

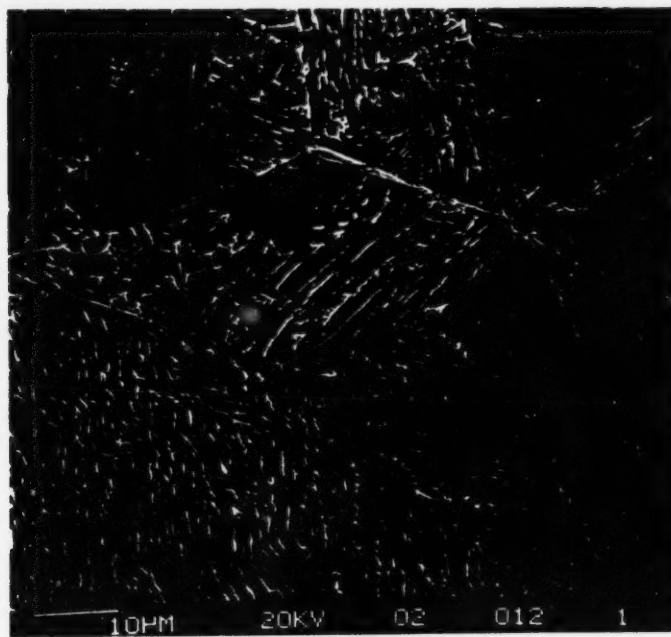


(b) SEM (neg.35 mm)

Fig. 6. Optical and SEM micrographs of 3 kJ/mm simulated CGHAZ microstructure in EH36 steel.

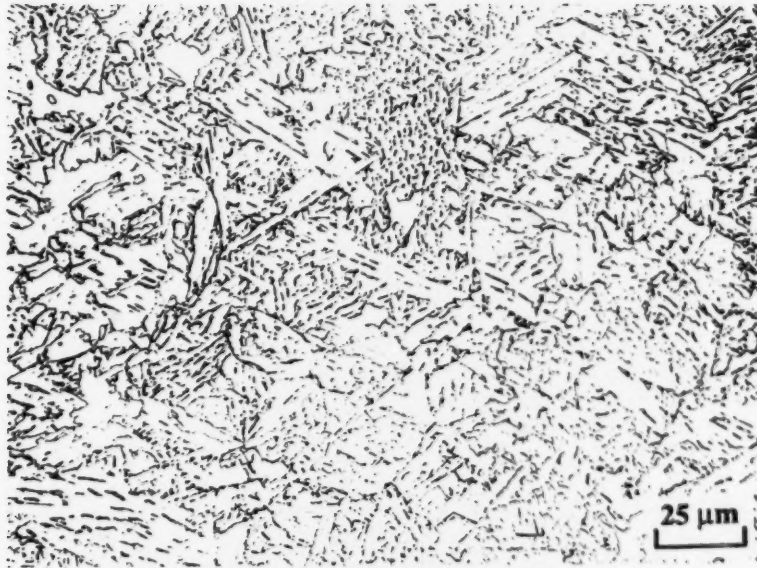


(a) Optical (neg.17343)

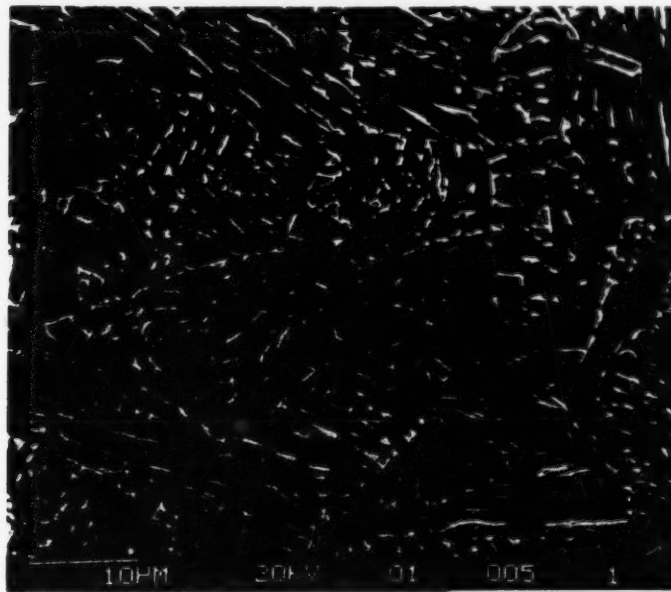


(b) SEM (neg.35 mm)

Fig. 7. Optical and SEM micrographs of 3 kJ/mm simulated CGHAZ microstructure in LT60 steel.

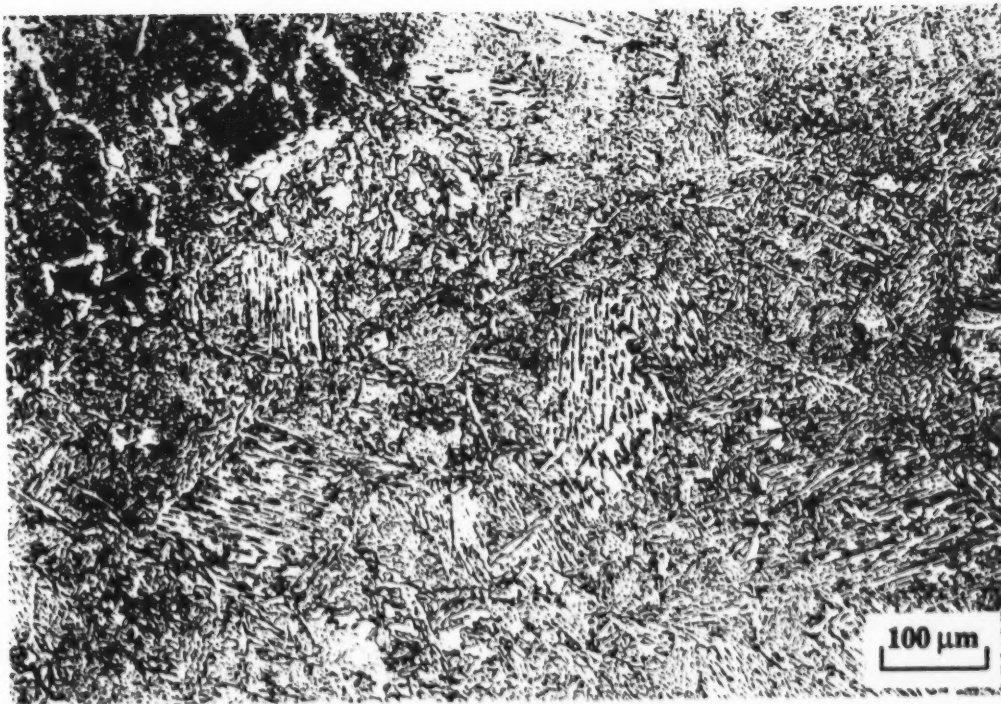


(a) Optical (neg.17344)

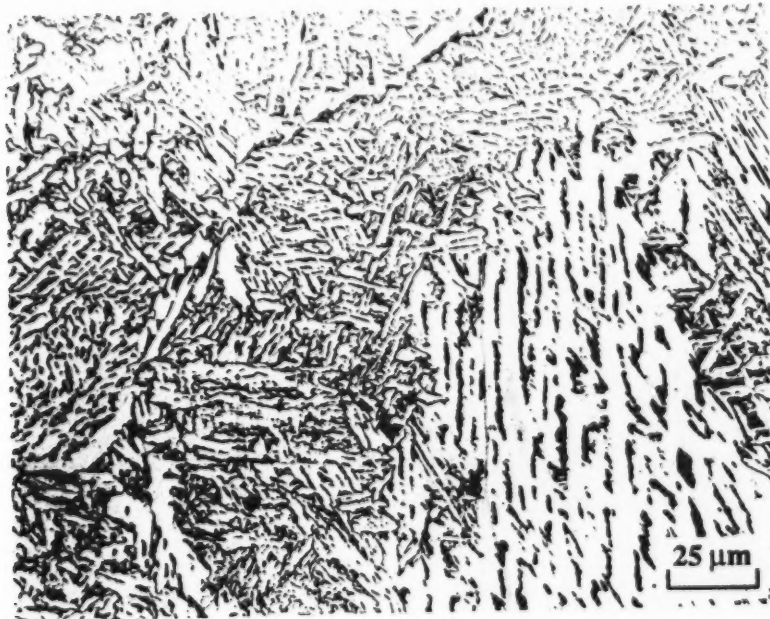


(b) SEM (neg.35 mm)

Fig. 8. Optical and SEM micrographs of 3 kJ/mm simulated CGHAZ microstructure in 350WI strand cast steel.



(a) (neg.17345)

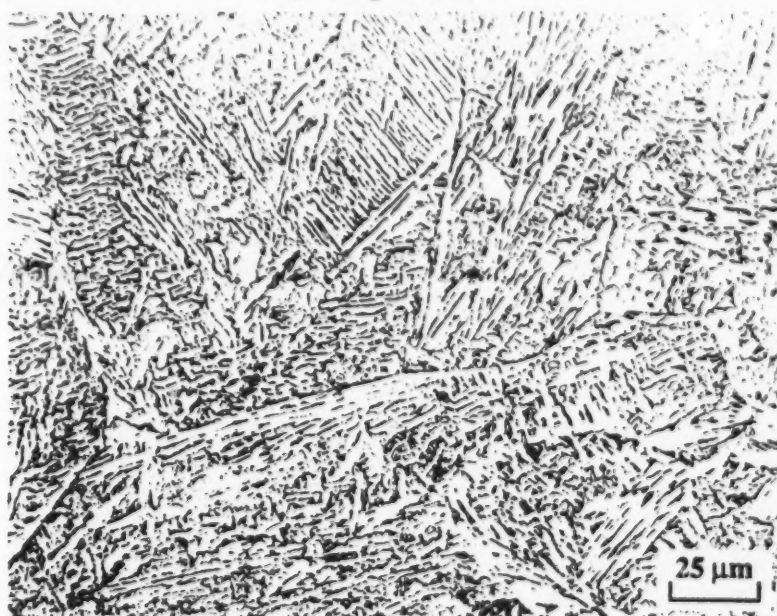


(b) (neg.17346)

Fig. 9. CGHAZ microstructure adjacent to the fusion line of submerged arc weld deposited at a 3 kJ/mm heat input in 350WT CAB steel. Weld MM 1.

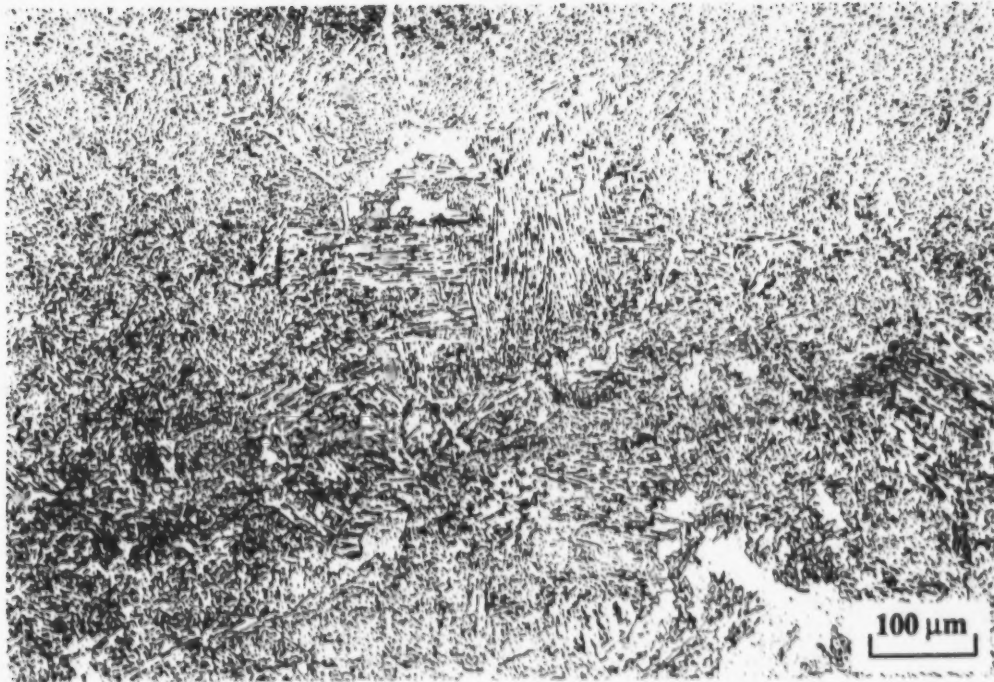


(a) (neg.17347)

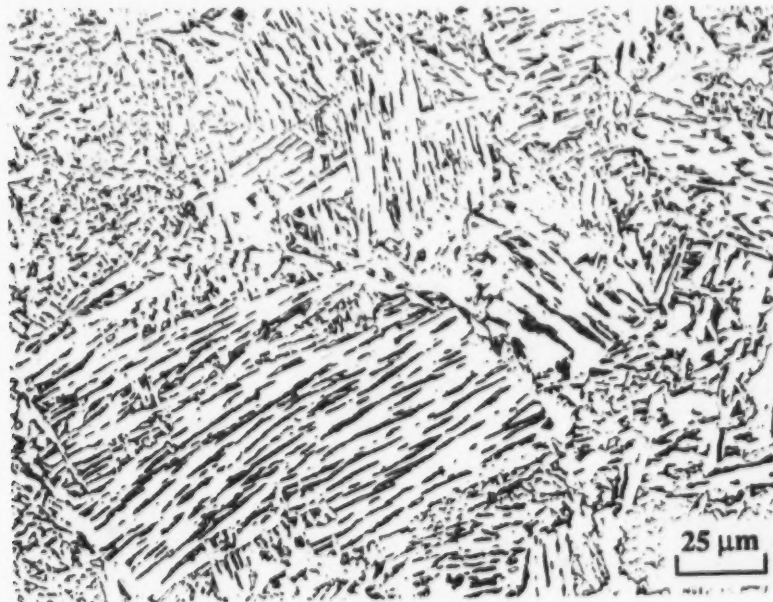


(b) (neg.17348)

Fig. 10. CGHAZ microstructure adjacent to the fusion line of submerged-arc weld deposited at a 3 kJ/mm heat input in EH36 steel. Weld MM 2.

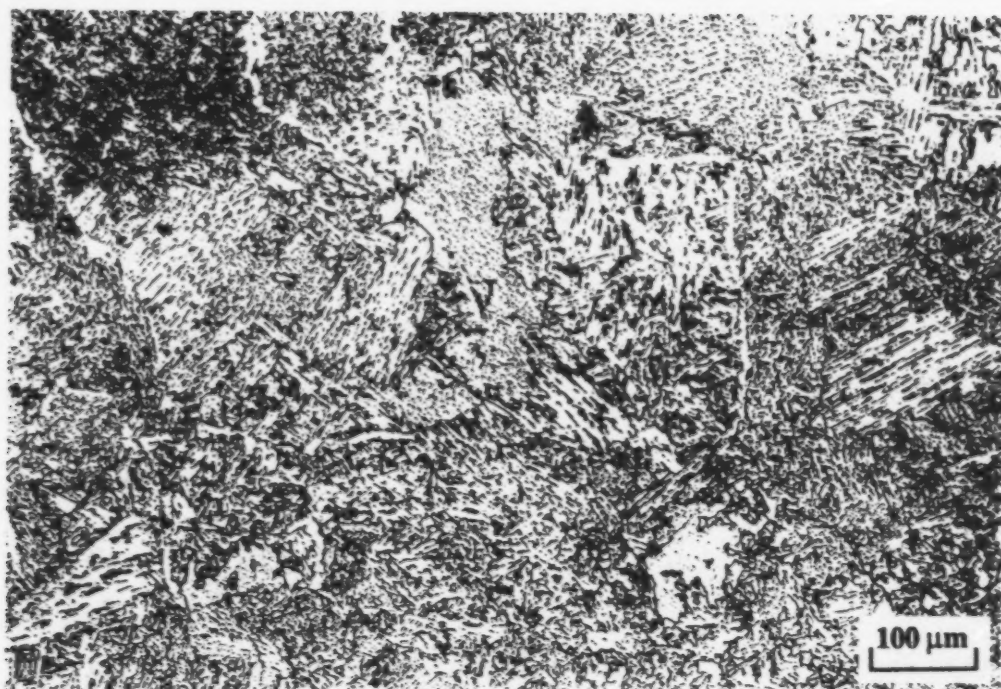


(a) (neg.17349)

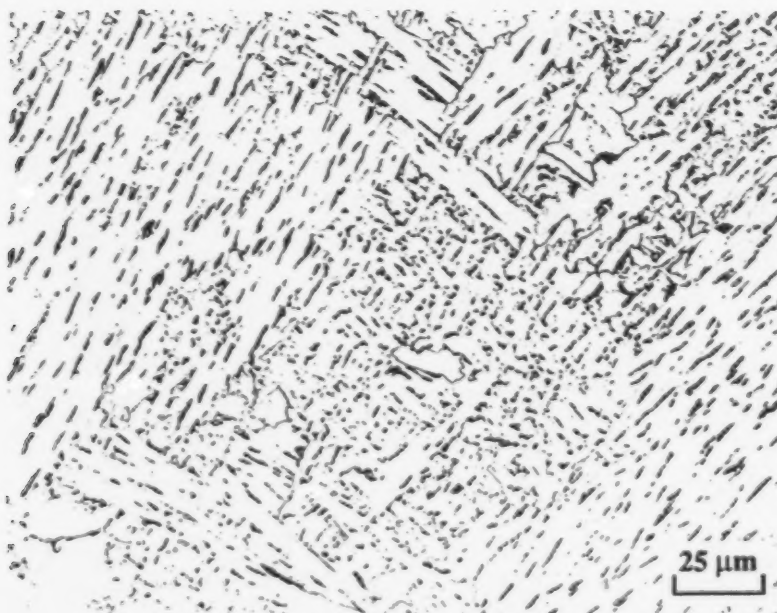


(b) (neg.17350)

Fig. 11. CGHAZ microstructure adjacent to the fusion line of submerged-arc weld deposited at a 3 kJ/mm heat input in LT60 steel. Weld MM 3.



(a) (neg.17351)

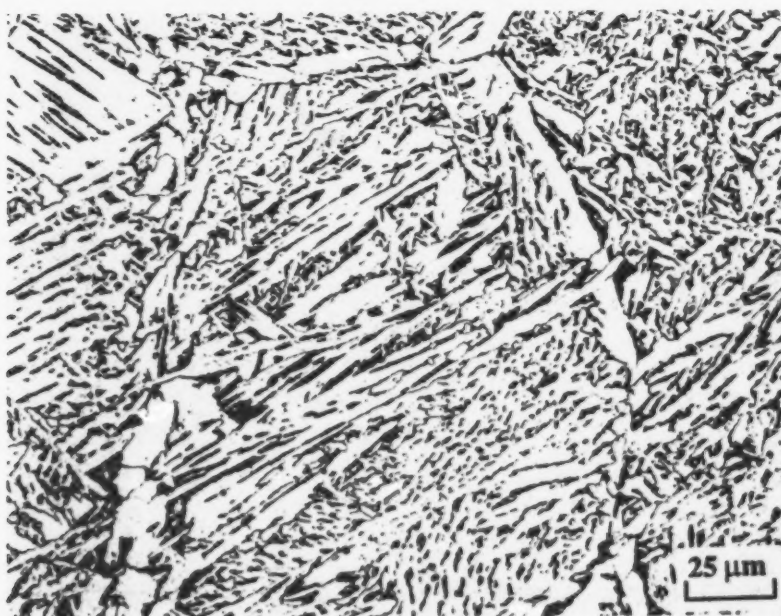


(b) (neg.17352)

Fig. 12 CGHAZ microstructure adjacent to the fusion line of submerged arc weld deposited at a 3 kJ/mm heat input in 350WT strand cast steel. Weld MM 4.

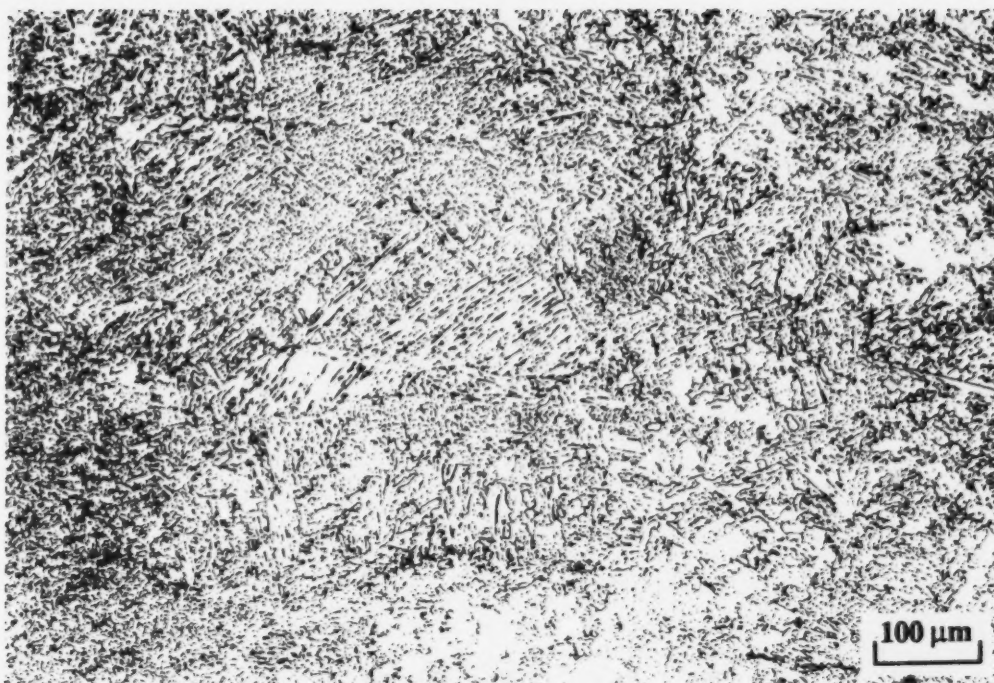


(a) (neg.17353)



(b) (neg.17354)

Fig. 13 CGHAZ microstructure adjacent to the fusion line of submerged arc weld deposited at a 6 kJ/mm heat input in 350WT CAB steel. Weld MM 5.

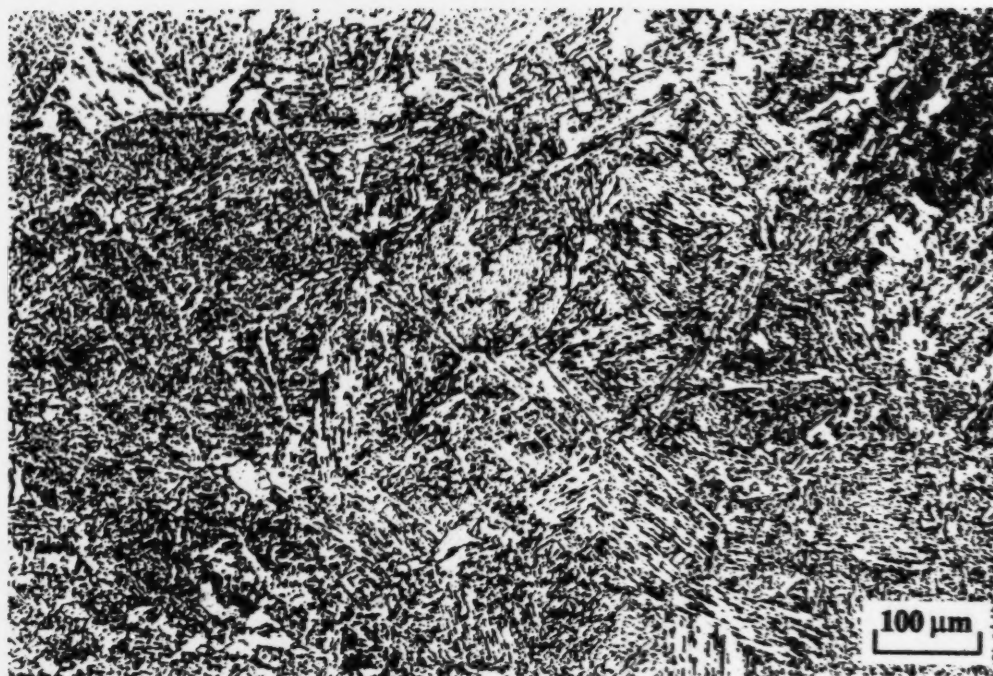


(a) (neg.17355)

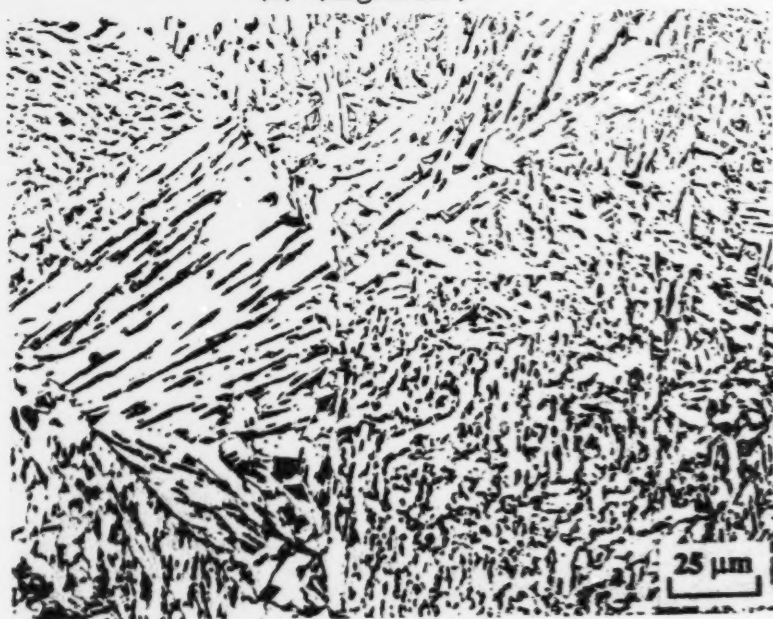


(b) (neg.17356)

Fig. 14. CGHAZ microstructure adjacent to the fusion line of submerged arc weld deposited at a 6 kJ/mm heat input in EH36 steel. Weld MM 6.

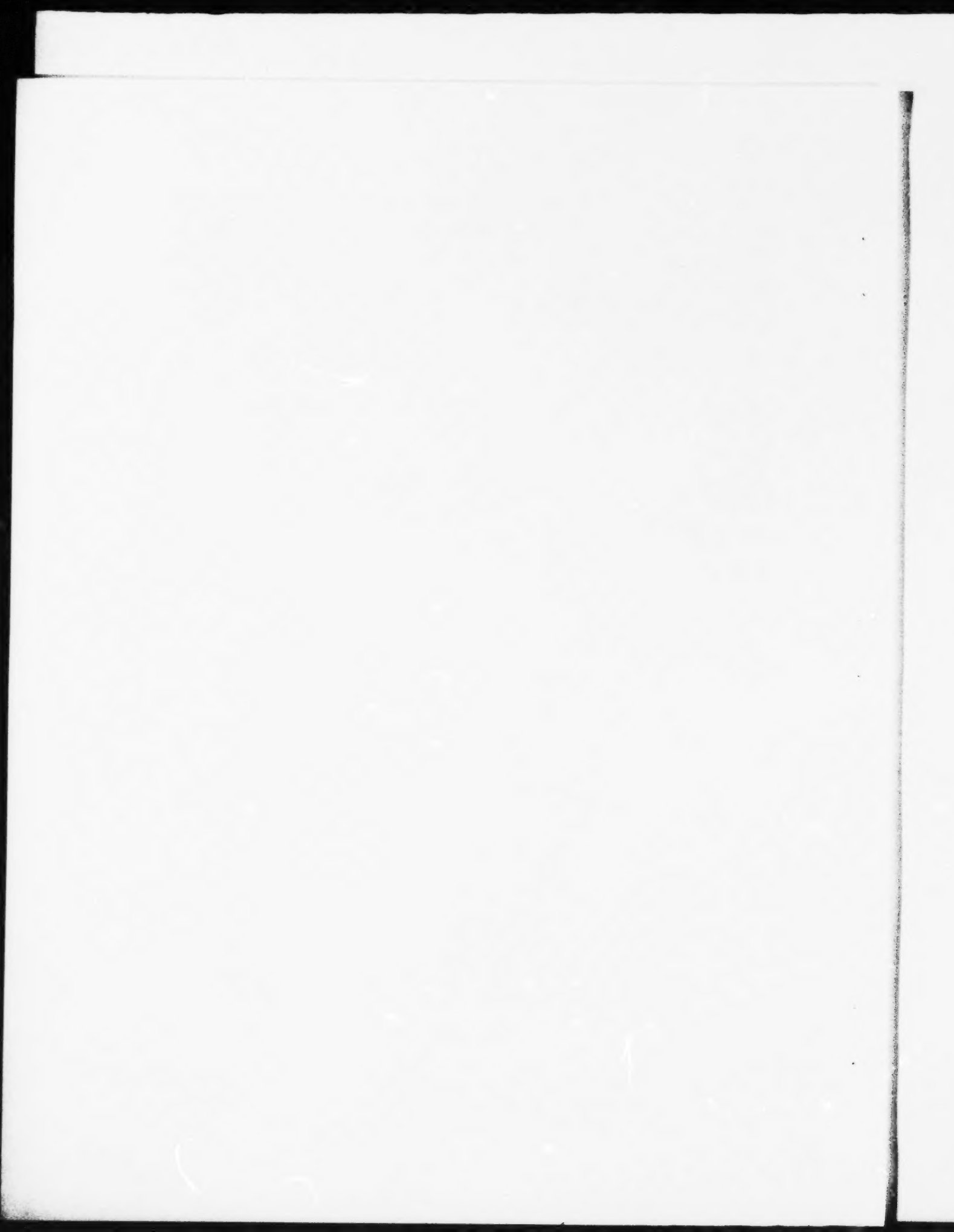


(a) (neg.17357)



(b) (neg.17358)

Fig. 15. CGHAZ microstructure adjacent to the fusion line of submerged arc weld deposited at a 6 kJ/mm heat input in LT60 steel. Weld MM 7.



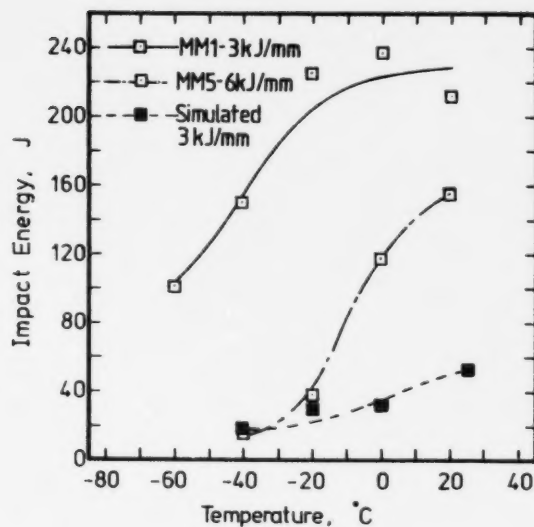


Fig. 17. Charpy transition curves of the CGHAZ of 350WT CAB steel (neg.15480).

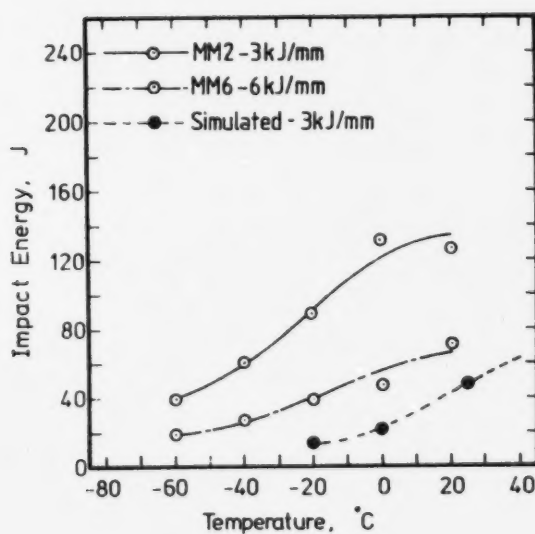


Fig. 18. Charpy transition curves of the CGHAZ of EH36 steel (neg.15481).

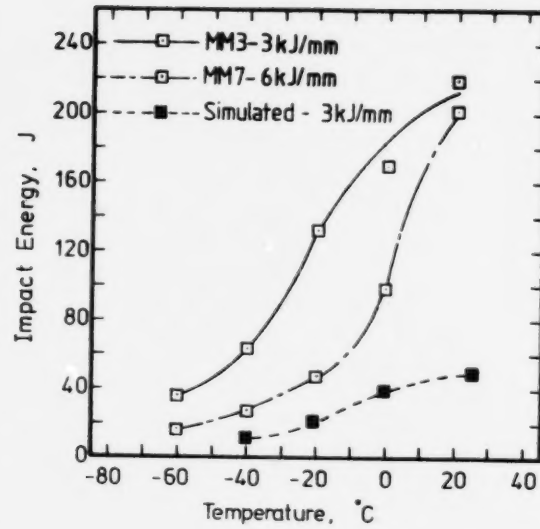


Fig. 19. Charpy transition curves of the CGHAZ of LT60 steel (neg.15478).

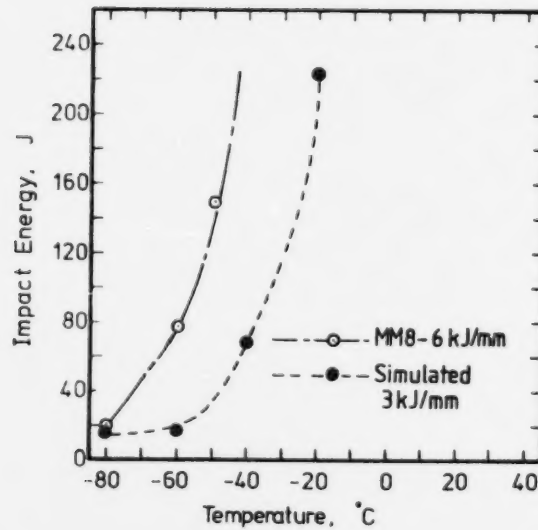
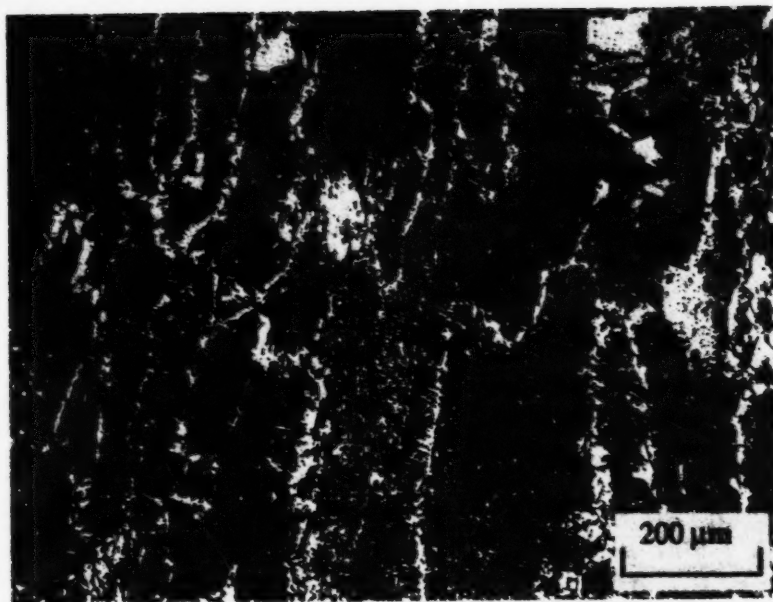
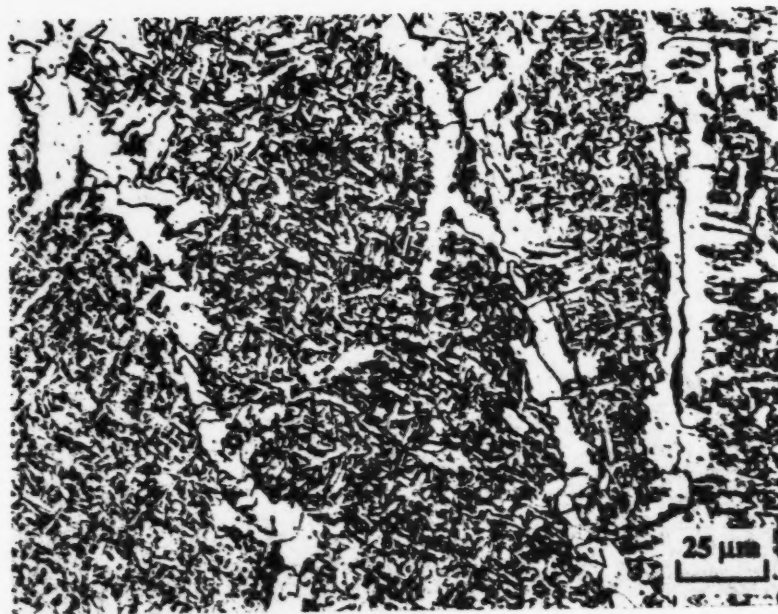


Fig. 20. Charpy transition curves of the CGHAZ of 350WT strand cast steel (neg.15479).

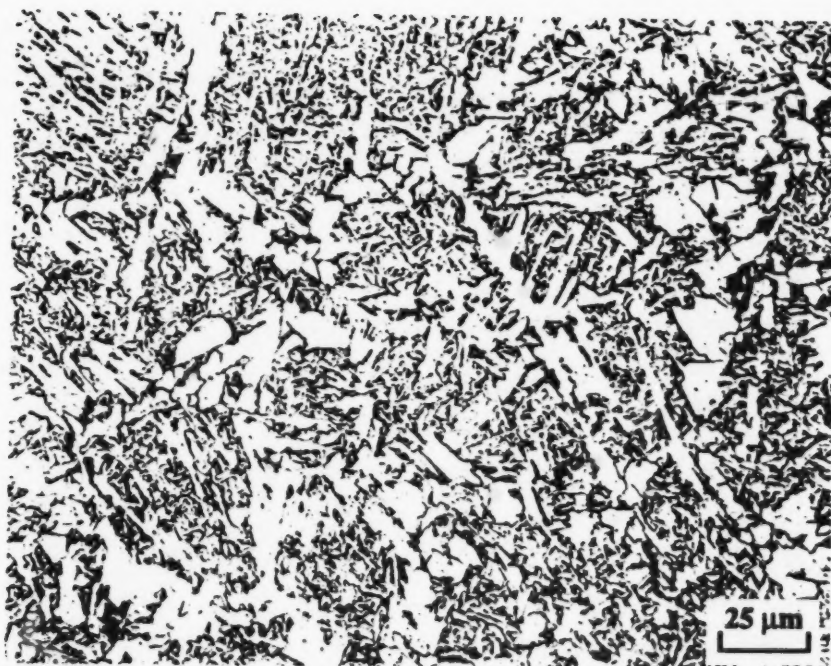


(a) columnar structure. (neg.17361)

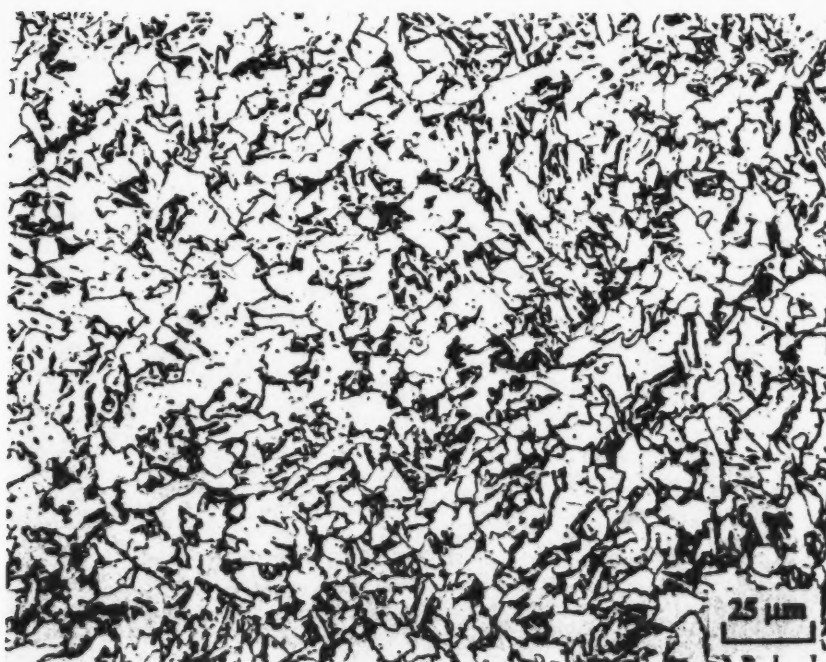


(b) microstructural constituents. (neg.17362)

Fig. 21. Microstructure of as-deposited region of submerged-arc weld deposited at 3 kJ/mm heat input. Weld M1.



(a) coarse grained reheated (neg.17363)

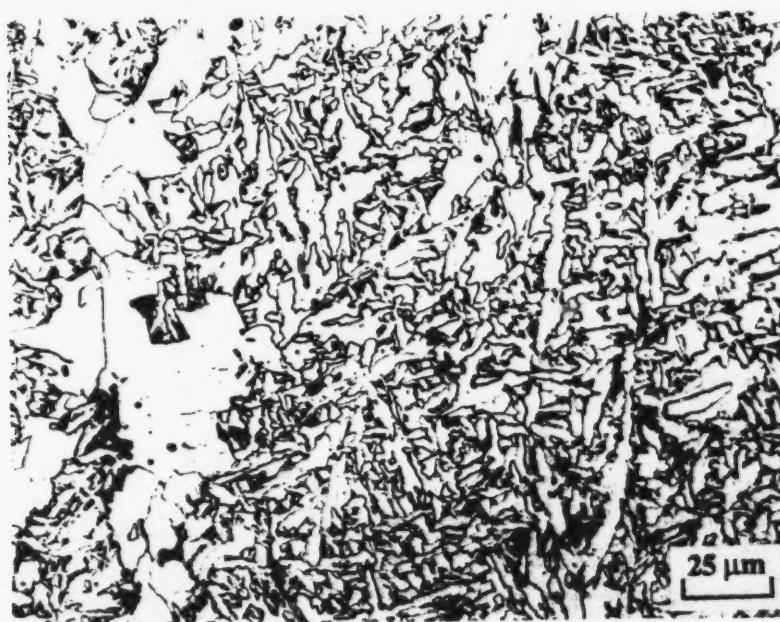


(b) fine grained reheated (neg.17364)

Fig. 22. Reheated weld metal microstructures of submerged-arc weld deposited at 3 kJ/mm. Weld M1.

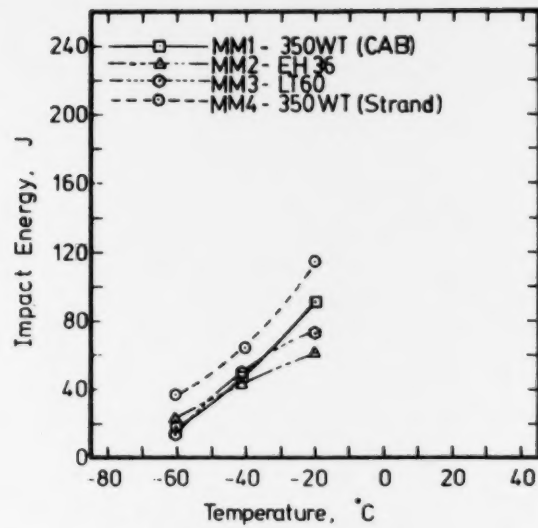


(a) columnar structure, (neg.17365)

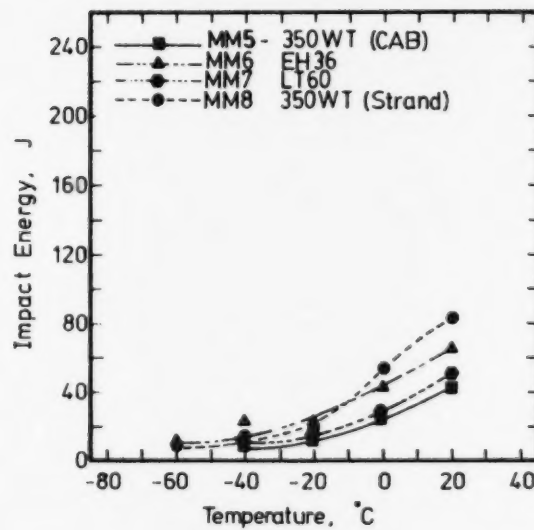


(b) microstructural constituents, (neg.17366)

Fig. 23. Typical weld metal microstructure of 6 kJ/mm submerged-arc welds. Weld MM8.



(a) 3 kJ/mm (neg.15483)



(b) 6 kJ/mm (neg.15482)

Fig. 24. Charpy impact transition curves for submerged arc welds.



(a) LT60 steel (neg.35 mm)



(b) 350WT strand cast steel (neg.35 mm)

Fig. 25. SEM fractographs showing cleavage facets of fractured Charpy specimens for the 3 kJ/mm simulated CGHAZ.

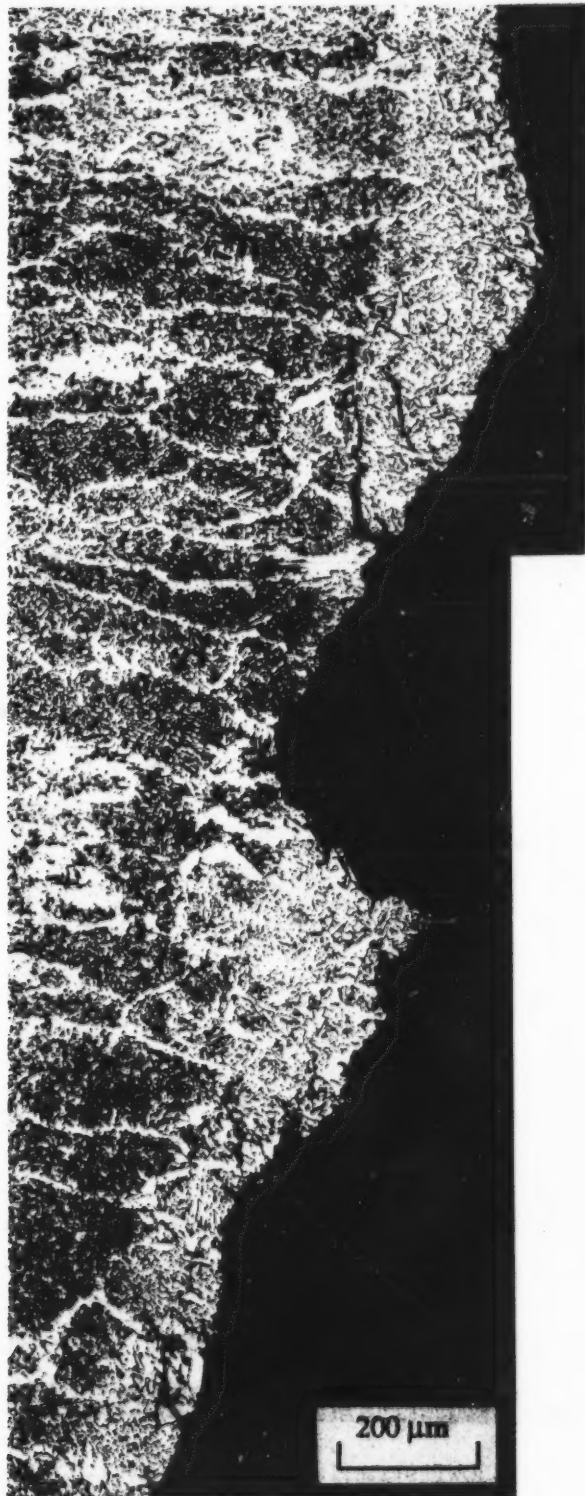


Fig. 26. Fracture path in 6 kJ/mm real weld HAZ specimen notched 1mm from the fusion line (negs.17367 and 17368).